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Carbon fibre composites with ceramic and glass matrices

Part 1 *Discontinuous fibres*

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The fabrication, microstructure and some of the mechanical and thermal properties of a series of composites are described. The systems investigated were magnesia, alumina, soda-lime glass, borosilicate glass and a lithia alumino-silicate glass-ceramic incorporating high modulus, chopped carbon fibres and magnesia containing chopped, stabilized zirconia fibres. Fracture strengths were increased when the fibres were partially aligned, but decreased when the fibres were randomly oriented. In all cases, however, a substantial increase in work of fracture was observed compared to the non-reinforced matrices. The observed effects are discussed in terms of the volume fraction of fibre, the mismatch of thermal expansion coefficients between matrix and fibre and the nature of the interface.

1. Introduction

Ceramics, including glasses and glass-ceramics, form a technologically important class of materials. Their principal attributes are their inertness and stability at high temperatures combined with their low density. Their theoretical strengths over the whole range of temperatures of technological interest compare favourably with the low temperature strengths of the strongest materials in common use but, in practice, bulk ceramics tend to exhibit low tensile strengths and poor resistance to thermal and mechanical shock. As a consequence, their use in structural applications is limited largely to situations where mainly compressive stresses are likely to be encountered.

The brittle behaviour of ceramics and the underlying causes have been discussed extensively in the literature. It occurs because ceramics do not undergo extensive plastic deformation prior to fracture and is reflected in the fact that the work of fracture, which is a measure of toughness, is for most ceramics of the order of 10 Jm^{-2}. For comparison, ductile metals exhibit works of fracture in the range 10^4 to 10^6 Jm^{-2}.

Considerable attention in recent years has been paid to the reinforcement of materials with fibres and whiskers and, under suitable conditions, increases in both strength and fracture energy can result. This paper describes some results of an experimental investigation into the conditions required to achieve improvement in the mechanical properties of ceramics by the incorporation of brittle, high strength fibres.

Six fibre-matrix systems were studied: carbon fibres in magnesia, alumina, borosilicate glass (Pyrex), soda-lime glass, and a lithia alumino-silicate glass-ceramic. The sixth system was zirconia fibres in magnesia and in addition, a zirconia powder-magnesia composite was investigated.

2. Fabrication

The composites were all produced by hot-pressing mixtures of fibres and powdered matrix material. High modulus type I carbon fibres were obtained from the Process Technology Division at AERE, Harwell in the form of long tows containing about 10^4 fibres of average diameter 8 μm, and were cut to convenient short

TABLE I Hot-pressing conditions for the composites.

System		Fibre loading vol %	Matrix powder size μm	Pressing conditions	
				Temperature °C	Pressure MNm^{-2}
Zirconia fibres in MgO	ZM	10, 20	< 1	Hot-pressed at 1000°C. Fired at 1200°C	6.9
Zirconia powder in MgO	ZPM	—			
Carbon fibres in MgO	CM	5, 15, 20, 40			
Carbon fibres in Al_2O_3	CA	10, 20, 30	~ 0.3 (Linde A)	1400°C	20.7
Carbon fibres in glass-ceramic	CGC	20	< 50	1230°C	10.4
Carbon fibres in borosilicate glass	CP	10, 20, 30, 40 (random)	< 50 (Pyrex)	700 to 1000°C depending on fibre content	6.9
Carbon fibres in borosilicate glass	CP	10, 20, 30, 40 (aligned)			
Carbon fibres in soda-lime glass	CS	5	< 50	700°C	3.5

lengths. The zirconia fibres were composed of the cubic phase stabilized with calcia and were produced by the H. I. Thompson Fibre-Glass Co, Gadena, California as a tangled wool of fibres whose individual lengths were a few centimetres and whose average diameter was 4 μm.

Powdered matrix material was made into a slurry with isopropyl alcohol, with the chopped carbon fibres or small pieces of zirconia wool added to the slurry during continuous blending with an Atomix high speed laboratory blender. In this way the zirconia fibres were separated and broken into short lengths typically about 200 μm long. Carbon fibres of initial lengths up to 3 mm could be dispersed easily to form a homogeneous mixture, but uniform dispersion became progressively more difficult as the fibre lengths increased, and above 10 mm interaction between fibres caused them to lie side by side in discrete bundles. For this reason a length of 3 mm was chosen for most of the carbon fibre mixtures. However, the initial cutting and the blending process of both zirconia and carbon fibres led to a wide scatter in the lengths of fibres in the final mixture and the mean length of carbon fibres in the pressed compacts was much less than 3 mm. The mixture was kept continuously agitated while excess isopropyl alcohol was removed by infra-red radiant heating until the mix was of a stiff consistency. In this condition it was loaded into a die assembly and hot-pressed.

Typical hot-pressing schedules are shown in Table I. For brevity, the systems studied are henceforth referred to by the code letters shown in the table. Pressing was carried out in a vacuum or in argon using graphite dies. Some alignment of the fibres occurred during pressing so that their long axes tended to lie in the pressing plane, but within this plane they were oriented randomly. In addition to these random fibre composites, samples of borosilicate glass containing partially aligned chopped carbon fibres were produced by a modification of the mixing and pressing technique.

Fabrication of the magnesia matrices followed a technique for producing pore-free magnesia [1] in which submicron magnesia powder was mixed with 1 % LiF powder and hot-pressed. The resulting compact was fired in argon at 1200°C to remove residual LiF. This technique was used for the CM, ZM and ZPM systems. ZM composites with 10 and 20 vol %* of fibres were successfully produced, but an attempt to make a 40 vol % sample resulted in a friable compact. The ZPM compact contained 10 vol % of calcia-stabilized ZrO_2 powder of particle size between 5 and 40 μm.

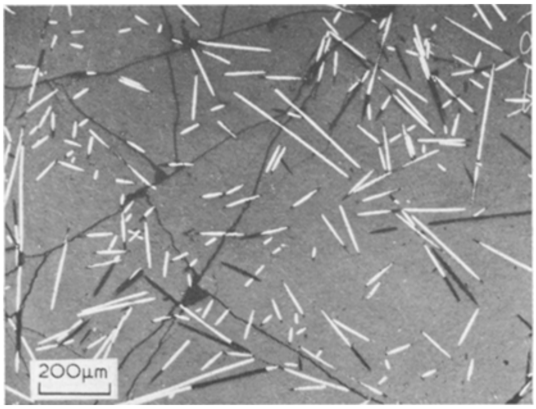

Figure 1 Polished section in the pressing plane of a 5 vol % CM specimen.

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*All fibre concentrations are expressed as volume percentages (vol %).

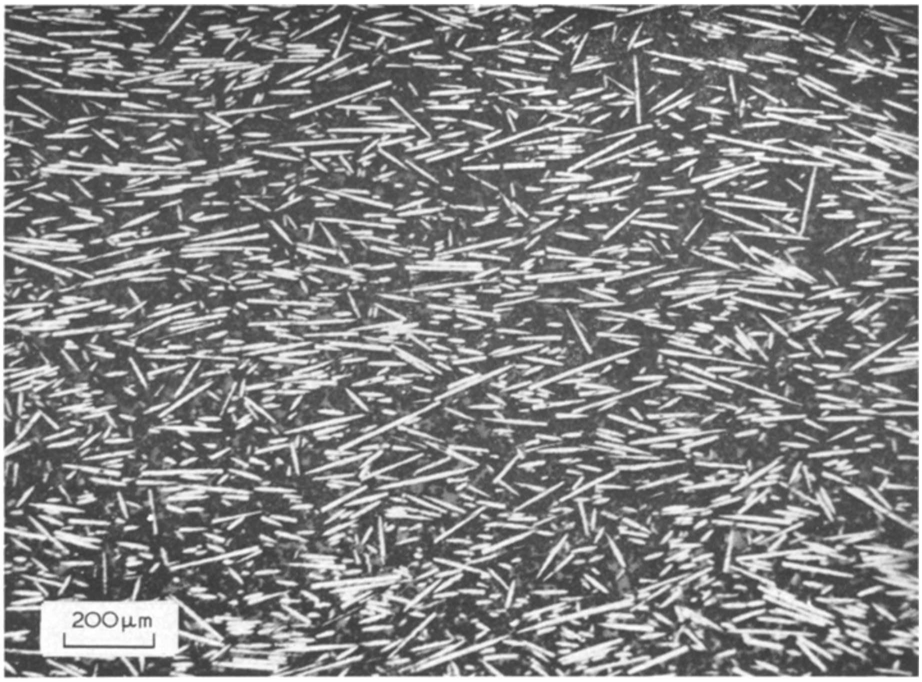

Figure 2 Polished section in the pressing plane of a 30 vol % aligned fibre CP specimen.

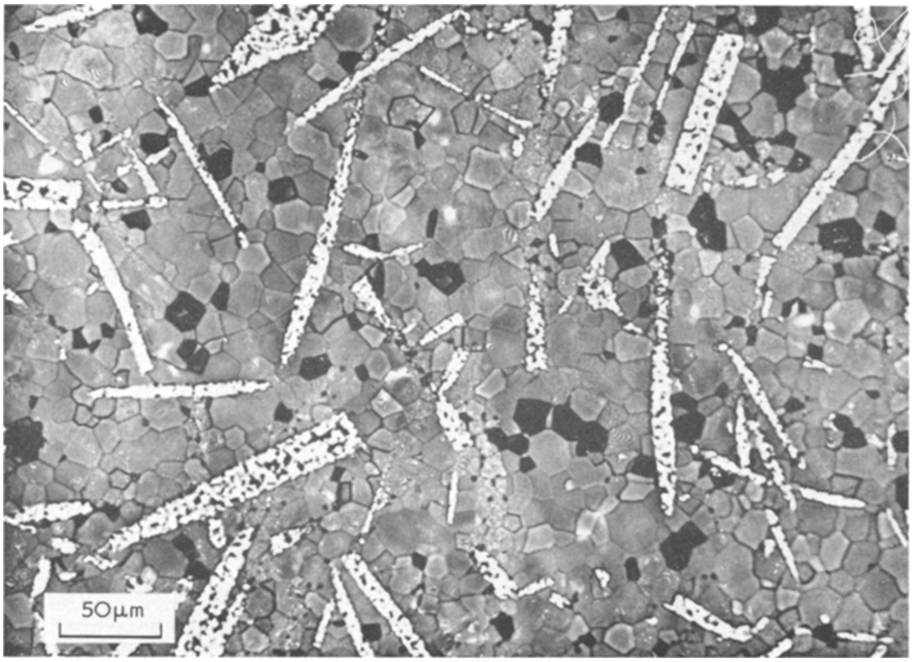

Figure 3 Polished and etched section in the pressing plane of a 10 vol % ZM specimen.

3. Microstructure

The microstructure of the composites was studied by optical microscopy of polished sections and by scanning electron microscopy. Figs. 1, 2 and 3 are polished sections showing the fibre distribution in the pressing plane for the 5 vol % CM, 30 vol % aligned CP and 10 vol % ZM specimens. Fig. 1 also reveals cracks in the matrix. Matrix cracking was characteristic of the CM, CA and CS composites and was most extensive in CM. None of the other systems displayed matrix cracking. Porosity was observed in the alumina matrix at all fibre concentrations and in the borosilicate glass at 40 vol %, but not in any of the other systems.

In all composites in the hot-pressed, untreated condition, unless extreme care was taken, mechanical polishing resulted in some degree of damage or "pluck-out" at the fibre-matrix interface suggesting that little or no chemical bonding had occurred. However, zirconia and magnesia react significantly at 1600° C and in an attempt to modify the bond, some of the ZM composite was heat-treated in air at that temperature for 16h. After this treatment, the polishing characteristics of the composite had changed and mechanical grinding damage at the fibre-matrix interface no longer occurred. After heat-treatment in air at 1700° C for 16 h the fibres were completely destroyed and the zirconia redistributed at the grain boundaries. Apart from an occasional giant grain, the presence of zirconia during these heat-treatments appeared to have inhibited grain growth in the magnesia. Similar behaviour was observed on heat-treatment of the magnesia composites containing zirconia particles.

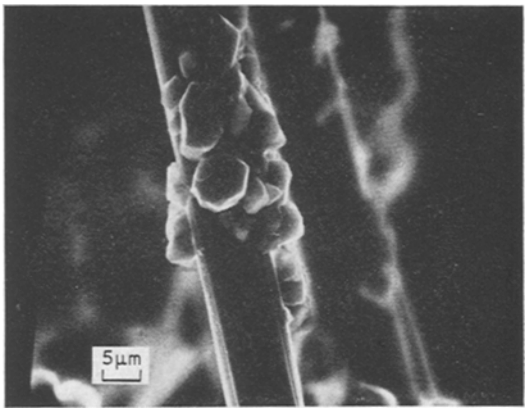

Figure 4 A carbon fibre with magnesia crystals adhering to it after pull-out during composite fracture.

The fracture surfaces of all the carbon fibre composites displayed considerable fibre pull-out, but in contrast there was no pull-out during fracture of ZM either before or after heat-treatment. In CM and CA, granular disintegration of the matrix occurred in the region of fracture, ceramic crystals adhered to the fibres as shown in Fig. 4, and the fracture surfaces of these matrices were rough and granular. In contrast the fracture surfaces of the matrices of CP and CS were smooth and conchoidal as shown in Fig. 5 and thin films of glass adhered to some of the pulled-out fibres. The fracture surface of CGC was more akin to that of the glass composites than the ceramic composites, but was less smooth because of the crystallinity of the matrix. A close examination of the pull-out holes in the glass and glass-ceramic composites revealed that the surface features of the fibres were replicated in the matrix, in a manner similar to that observed in carbon fibre-resin matrix composites, because of the fluidity of the matrix during hot-pressing. There was no evidence of such replication in the alumina and magnesia matrices.

4. Mechanical property measurements

Bend strengths and works of fracture were determined from three-point bend tests on specimens in the "as-cut" conditions using an Instron machine with an electromechanical integrator. The work of fracture was determined by fracturing a specimen in a controlled manner so that the energy required to create unit area of fracture face could be measured [2, 3] and has been taken as a measure of the fracture toughness of the composites. Specimens were in the form of rectangular prismatic bars and the neutral plane in bending was parallel to the hot-pressing plane of the compact. Aligned fibre composites were tested with the maximum tensile stress in the mean direction of the fibre long axes.

5. Mechanical property results

Typical load-deflection curves for unreinforced Pyrex and CP specimens with a length to depth ratio of 10 are shown in Fig. 6. The low fracture energy of glass resulted in catastrophic failure when the unreinforced sample reached its critical strain. Failure of the CP specimen was quite different, however, and propagation of the crack proceeded in a progressive and controlled manner with increasing displacement of the crosshead because of the higher fracture energy of the

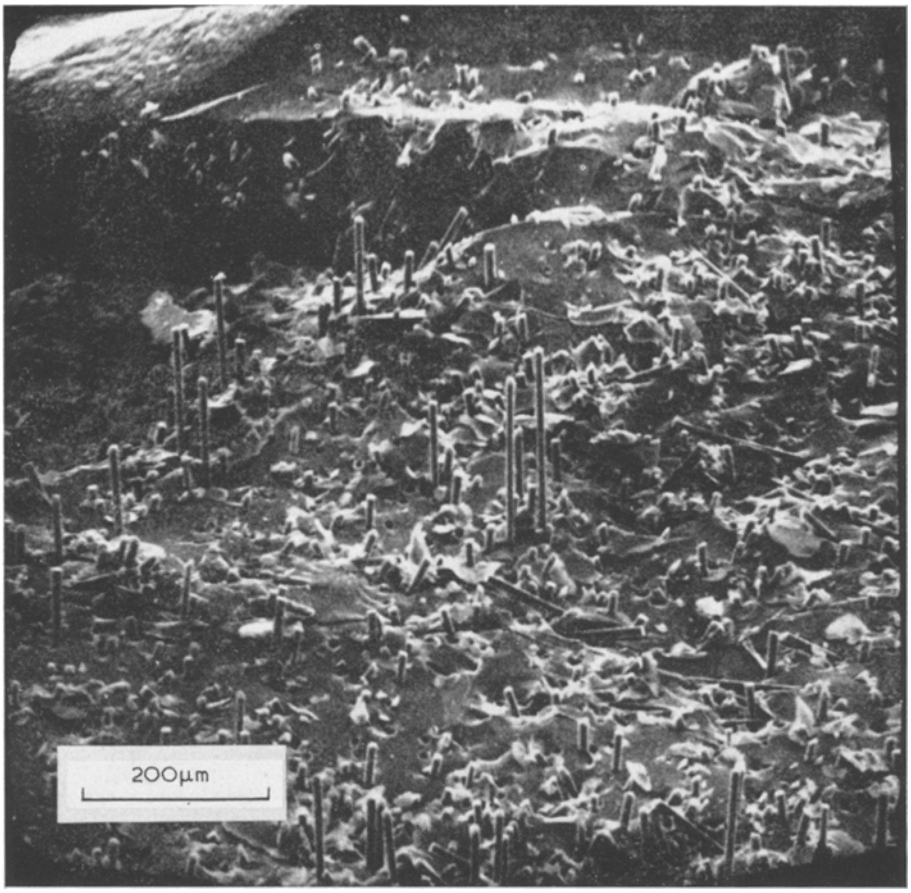

Figure 5 A fracture surface of a CP composite.

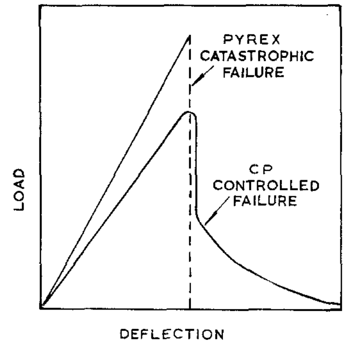

Figure 6 Load-deflection curves for unreinforced Pyrex and CP.

composite. Increased works of fracture were obtained in all the composite systems and the results of the measurements are shown in Fig. 7 as a function of fibre concentration. To obtain controlled failure of a low fracture energy material, and thus measure the work of fracture, it is necessary to notch the bend specimens. As the fracture energy increases it becomes unnecessary to notch. It was necessary to notch some of the composites studied, but not all, and the works of fracture have thus been measured off specimens with varied notch depths. Davidge and Tappin [3] have shown that notch depth can have an effect on the measured work of fracture values of non-fibrous ceramic material, but any such effect on composites has not been taken into account here.

The variation of the bend strength (σ_c) relative to that of the matrix (σ_m) is shown in Fig. 8 as a function of the fibre concentration. The values of σ_m quoted on the graph were determined on unreinforced samples pressed under conditions identical to those used to produce the composites. The introduction of

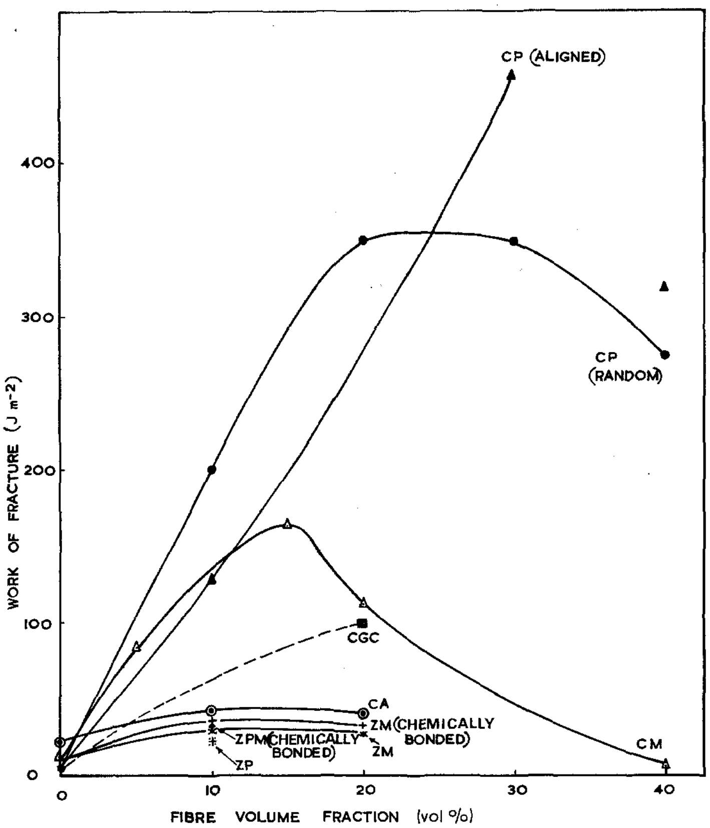

Figure 7 The works of fracture of the composite systems.

fibres resulted in a decrease in strength for all the randomly oriented composites but, after an initial decrease, the strength of the aligned fibre composite increased with fibre content up to 30 vol %. The drop in strength at 40 vol % was associated with the presence of large voids due to poor compaction of the matrix. In contrast with the behaviour of continuous fibre-resin matrix composites there was no indication of shear failure during bending in any of the systems studied, and failure always occurred by crack propagation from the tensile face.

The temperature dependence of strength of some of the composites and matrix materials is shown in Fig. 9 and the effect of temperature on the work of fracture of 20 vol % random fibre CP is shown in Fig. 10. In each of these tests, the time taken in heating the sample was about 5 min and measurements were made in air as soon as the test temperature was reached.

The thermal shock resistance of 10 vol % random fibre CP has been assessed by heating samples in a furnace to a known temperature and then transferring them rapidly to a water bath. The extent of the resulting damage was determined by revealing microcracks with a dye penetrant technique, using a dye sensitive to ultra-violet light, and by measurement of the specimen strength in three-point bending at room temperature. Specimen dimensions were 14 × 4.5 × 4.5 mm. Fig. 11 shows the variation of bend strength with quench temperature. There was no indication of surface cracking on specimens which had been quenched at 100° C intervals up to 400° C, but on quenching from 500° C a mosaic of very fine, shallow microcracks appeared. Above 600° C there was an overall crazing of the surfaces without deep penetration of the cracks. Thus, on quenching, the decrease in strength of CP appears to be associated with the onset of surface cracking. This behaviour differs from that of Pyrex in that cracking of the

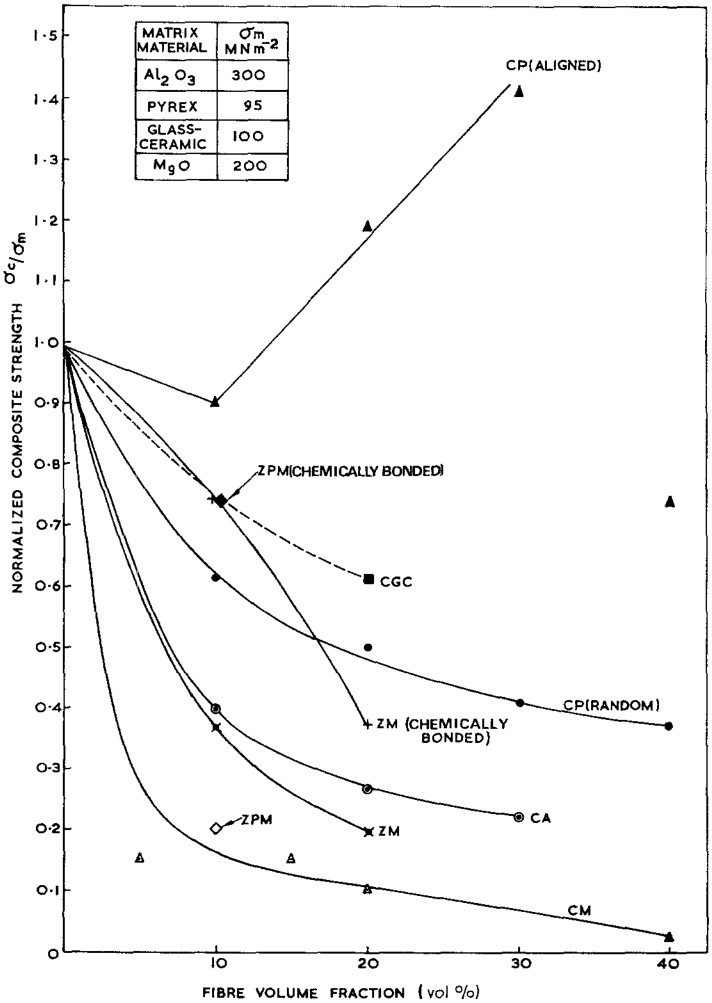

Figure 8 The variation of bend strengths with fibre concentration.

unreinforced material can be detected on quenching from temperatures in excess of 275° C and is considerably more severe [4].

6. Discussion

The mechanical properties of fibre composites are determined, to a large extent, by the strength and nature of the bond between the fibres and the matrix. In the case of the carbon fibre composites considered in the present work, it does not appear that any significant degree of chemical bonding exists at the interface, but that the properties depend on the development of a mechanical bond determined by differences in the thermal expansion coefficients of the fibres and the matrix and by interlocking or keying of the matrix with irregularities in the surface of the fibres. The properties of these composites are further affected by the occurrence of matrix cracking due to the mismatch of thermal expansion between fibres and matrix.

The importance of the thermal expansion coefficients arises from the fact that the composites were fabricated at high temperature. As the composite is cooled after hot-pressing, it reaches a temperature T_c below which stress-relaxation processes in the matrix are halted and on further cooling, stresses are built up in the matrix and fibres due to thermal mismatch. In considering these stresses, it is necessary to take into account the anisotropy in the thermal expansivity of carbon fibres [5], the coefficient (α_a) being small

TABLE II Matrix properties and calculated thermal stress parameters and ratios.

Matrix	α_m (°C^{-1}) × 10^6	T_c (°C)	E_m (GNm^{-2})	σ_m MNm^{-2}	σ_a MNm^{-2}	σ_r MNm^{-2}	ϕ_a	ϕ_r
MgO	13.6	1200	300	200	4900	2020	25	10
Al_2O_3 (80% dense)	8.3	1400	230	300	2660	97	9	0.3
Soda-lime glass	8.9	480	60	100	260	26	2.6	0.3
Borosilicate glass	3.5	520	60	100	110	− 140	1.1	− 1.4
Glass-ceramic	1.5	1000	100	100	150	− 650	1.5	− 6.5

Type I carbon fibres $\alpha_a \sim 0$, $\alpha_r \sim 8 \times 10^{-6}$ °C^{-1}.

in the axial direction and relatively large in the radial direction (α_r) (Table II).

The stress in the matrix due to mismatch in the axial direction is primarily a function of the parameter $\sigma_a = (\bar{\alpha}_m - \bar{\alpha}_a) \Delta T E_m$ where ΔT is the temperature decrease below T_c, $\bar{\alpha}_m$ and $\bar{\alpha}_a$ are the mean expansion coefficients of the matrix and fibre respectively over that temperature range, and E_m is Young's Modulus of the matrix at the final temperature. If this stress parameter is positive it implies that the matrix is in tension and increased values of the parameter result in increased stresses in the matrix. In the radial direction the stress is similarly primarily a function of the parameter $\sigma_r = (\bar{\alpha}_m - \bar{\alpha}_r) \Delta T E_m$ which when positive implies that the matrix is in tension due to hoop stresses and when negative implies that the fibres tend to shrink away from the matrix. Matrix cracking occurs when the local tensile stresses due to thermal mismatch are greater than the matrix tensile strength σ_m. Thus when the stress ratios $\phi_a = \sigma_a/\sigma_m$ or $\phi_r = \sigma_r/\sigma_m$ reach critical positive values the matrix will crack. The exact functional relationship of the thermal stresses to $(\alpha_m - \alpha) \Delta T E_m$ depends upon the interaction of stresses from all the fibres and thus depends upon the geometry of the fibre array, fibre dimensions, elasticity of fibres and matrix, fibre-matrix bonding, and on local stress concentrations. In the absence of an exact evaluation of this relationship only qualitative observations can be made using the calculated thermal stress parameters. In the case of the ceramic matrices, it is unlikely that appreciable

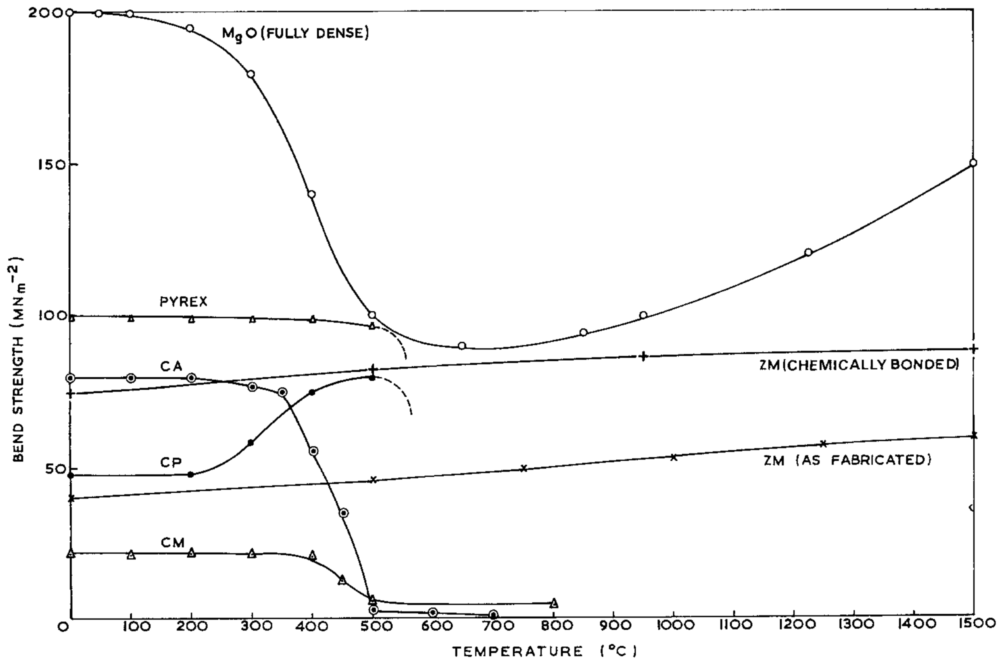

Figure 9 The temperature dependence of strength in air of some composite systems and matrices.

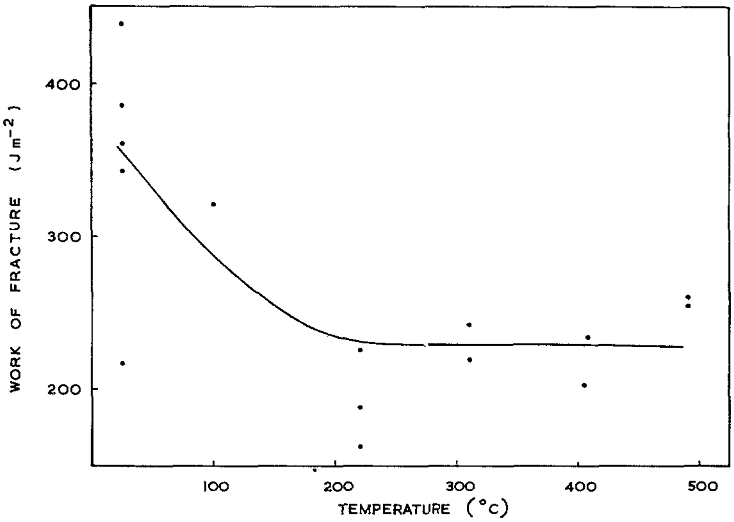

Figure 10 The temperature dependence of the work of fracture in air of 20 vol % random fibre CP.

stress relaxation occurs by plastic flow at the temperatures employed and T_c has been taken accordingly as the pressing temperature. In the glasses, on the other hand, stress relaxation could take place by viscous flow and T_c has been taken as the strain point. Table II shows the thermal stress parameters and the stress ratios, together with the values of expansion coefficients, elastic moduli and relaxation temperatures adopted for their calculation. With the exception of magnesia the radial stress ratios are either small or negative, the negative value indicating that the fibre shrinks away from the matrix during cooling. This component of stress would not be expected, therefore, to give rise to cracking in the matrix. The axial stress ratios however, are all positive and give rise to tensile stresses in the matrix parallel to the fibre axis. On this simple picture, cracks would tend to form perpendicular to the axes of the fibres and this seems to be borne out by Fig. 1. The degree of cracking should presumably be related to the magnitude of the stress ratios and should therefore follow the matrix sequence M, A, S, P, GC in descending order. No cracking was observed experimentally in the CP and CGC composites,

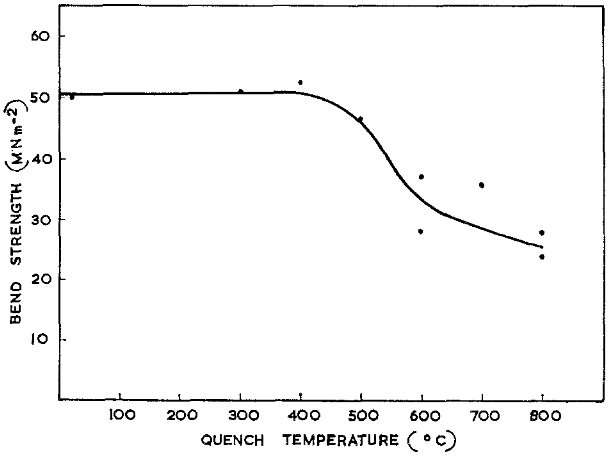

Figure 11 The effect of quench temperature on the bend strength of 10 vol % random fibre CP.

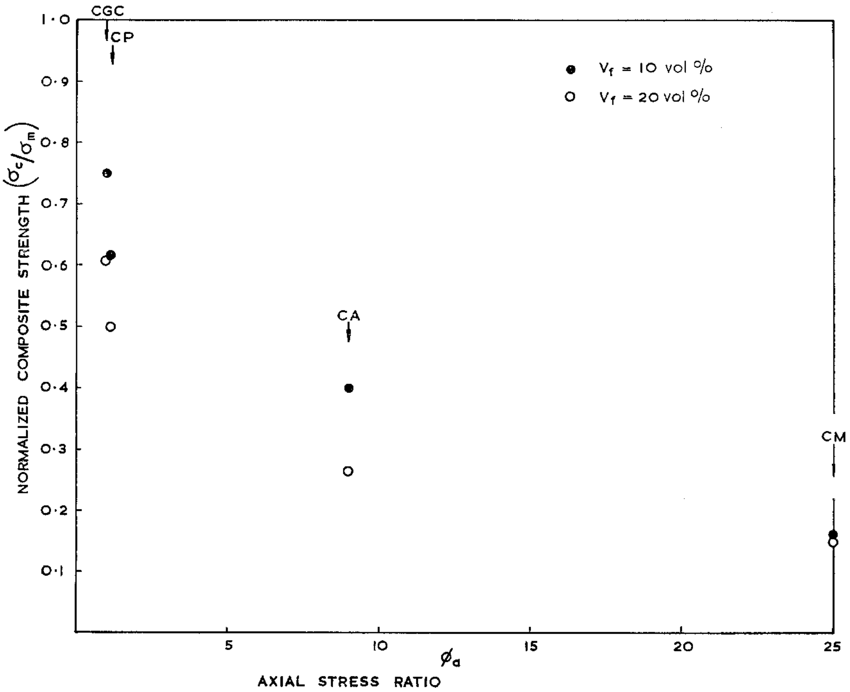

Figure 12 Normalized composite strength (σ_c/σ_m) plotted as a function of axial stress ratio (ϕ_a).

but the extent of cracking in the CM, CA and CS samples followed the predicted trend. The strengths of these composite materials appears to depend largely on the magnitude of the tensile stresses frozen into the matrix. This is demonstrated by Fig. 12 which shows the relative strengths σ_c/σ_m of the similar, random carbon fibre composites plotted against ϕ_a.

Apart from thermal stresses, another factor is responsible for the decrease in strength of the random fibre systems. The reinforcing effect of the fibres is diminished by their non-alignment, and those fibres that are not reinforcing weaken the composite by acting as stress concentrators in the brittle matrix. The influence of this effect can be seen by comparing the strengths of the aligned and random fibre CP samples (Fig. 8).

It is instructive to compare the strengths achieved by the "aligned" fibre composites with those expected on the basis of the simple mixtures theory [6]. Fig. 5 shows fibres protruding from the fracture face of a CP composite. The maximum length approaches the transfer length of the fibre-matrix combination and for this particular specimen $l_c \sim 200$ μm. On the basis of a mean fibre length of 3 mm, the strength of a perfectly aligned 30 vol % fibre composite with a typical carbon fibre strength of 2000 MNm^{-2} might be expected to approach the continuous carbon fibre composite strength of $\sim$ 600 MN/m^2. The low strengths observed experimentally are due to a combination of poor alignment and the breakage of fibres during fabrication into a distribution of lengths, many of which were less than l_c.

The effect of temperature on the bend strength, shown in Fig. 9, is clearly different in the case of the composite with an uncracked matrix, (CP), from those where matrix cracking occurred during fabrication (CA and CM). The strength of CP increased between 200 and 500° C and this can be ascribed principally to a reduction in the longitudinal thermal stress parameter, which, as the temperature approaches T_c, tends to zero. Above T_c, however, the matrix ceases to become a rigid elastic solid and begins to deform by viscous flow. Thus the strength can be expected to reach a maximum at a temperature close to the strain point (520° C). An additional strengthening effect is provided by the difference in radial expansion of the fibre and matrix. As the temperature is increased, the fibre expands so as to be gripped more tightly by the matrix, and this increased bonding is expected to lead to increased strength [6].

The variations in the strength of the ceramic composites CA and CM between 350 and 500° C are not believed to be due primarily to changes in matrix properties, in spite of the relatively large temperature dependence of strength for unreinforced magnesia. The latter is due to the increasing mobility of dislocations with temperature, but plastic flow is known not to occur in this temperature range in alumina. More probably, the extensive cracking in these matrices as a result of thermal stresses during fabrication reduces the contribution of the matrix to the strength of the composite to a very low level. Conversely, although the strength is reduced, the effect of the fibres in magnesia appears to be to reduce the sensitivity of the bend strength to temperature up to ~ 400° C. Above this temperature, it is believed that the fall in strength in both CM and CA is a result of progressive oxidation of the fibres. Bulk graphite begins to oxidize significantly in air above 400° C and compared to the CP and CGC samples, oxidation of the fibres is enhanced by the open paths for oxygen through the cracks in the matrix.

The strengths of the composites have this far been discussed principally in terms of the longitudinal thermal stress and its influence on the generation of cracks in the matrix. Transfer of stress from matrix to fibres however, depends on the degree of bonding and in the systems under consideration, this is mainly a function of the radial stress parameter. Table II shows that the fibres are gripped positively by the magnesia, alumina and soda-lime glass matrices, but that in the CP and CGC composites the fibres shrink away from the matrix on cooling below T_c. In these glasses, bonding could result from the elastic strain frozen into the composite, because pressure is maintained during fabrication until the matrix is rigid; and from mechanical keying due to the matrix moulding to irregularities in the fibre surface. The radial displacement of the fibres from the matrix would be ~ 100Å for CP and ~ 250Å for CGC on cooling from their respective relaxation temperatures. Convolutions greater than this can be detected on fibre surfaces and thermal shrinkage would not therefore be expected to completely remove the bond. An order of magnitude calculation shows that the contribution of the frozen-in elastic strain to the bonding is negligible. Thus the available evidence supports mechanical keying as the principal mechanism of bonding in CP and CGC, while the lack of replication of the fibres in the M and A matrices suggests that thermal shrinkage of the matrix provides the necessary bonding. A combination of the two occurs in the CS system.

TABLE III Thermal strain parameters (at 20° C) and measured works of fracture for 20 vol % carbon fibre composites.

Material	Work of fracture Jm^{-2}	$(\bar{\alpha}_m - \alpha_r)\,\Delta T$
MgO	10	
CM	110	6.6×10^{-3}
Al_2O_3	38-66	
CA	40	0.4×10^{-3}
Pyrex	4	
CP	344	-2.3×10^{-3}
Glass-ceramic	4	
CGC	100	-6.0×10^{-3}

The work of fracture of all the systems increases with the addition of fibre. Increased work of fracture of fibre composites can result from several mechanisms. A propagating crack can be deflected by weak interfaces and delamination of an aligned fibre composite can occur due to poor tensile strength perpendicular to the axis of the fibres [7]. Pull-out of fibres from fracture faces dissipates energy and can result in high works of fracture if the fibre-matrix shear strength is maintained during pull-out. The fracture face shown in Fig. 5 is typical of the carbon fibre systems studied in that there is no evidence of delamination, or of the creation of a significant amount of fracture surface by deviation of the crack at the interfaces between the fibres and matrix. The appearance of the fracture faces suggests that the increased work of fracture is due to fibre pull-out. Table III compares the room temperature work of fracture measurements of the 20 vol % random carbon fibre systems with the radial, thermal strain parameters $(\bar{\alpha}_m - \alpha_r)\,\Delta T$. The works of fracture of the matrices are negligible compared with the works of fracture of the CP and CGC composites and, since the matrices of CM and CA are badly cracked prior to fracture, it is not unreasonable to assume that the work of fracture contribution from the matrix in these materials may be ignored. Thus the major contribution to the work of fracture is expected to result from pull-out. Reference to Table III indicates that, if the ceramic (CM and CA) and glass (CP and CGC) matrix systems are considered as separate groups, the more tightly bound material within each group (CM and CP respectively) has the higher work of fracture. It is appropriate to consider these materials in two groups because of the difference in keying that can occur during fabrication of the two types of matrix. The work of fracture of a fibre composite due to fibre pull-out is inversely dependent on the shear strength of the interface [6]. On analogy with resin systems it might therefore be expected that the work of fracture of the tightly gripped fibre systems would be less than the less tightly bound systems. Table III shows that this does not appear to be the case. On the other hand, the temperature dependence of the work of fracture of CP (Fig. 10) suggests that the work of fracture decreases as the mechanical bonding improves, since an increase in temperature would reduce the relative shrinkage of the fibre away from the matrix. An analysis of the influence of bonding on the work of fracture clearly requires more detailed information on the distribution of fibre lengths and critical transfer lengths than has been obtained in the present work.

Inclusion of zirconia fibres into a magnesia matrix would not be expected to produce a stronger material than the matrix alone because the modulus of the fibres is lower than that of the magnesia. The interest in this system lies in the ease with which the fibre-matrix bond may be altered. As fabricated, ZM is mechanically bonded. However, this may be altered to a partly chemical bond by suitable heat treatment and Fig. 8 shows that the improved bonding results in an improved strength. The expansion coefficients of the zirconia fibres are the same in radial and axial directions and are less than those of the magnesia matrix, so that at room temperature the matrix is in tension. On increasing the temperature these tensile stresses are progressively relieved and this is believed to be the explanation of the increase in strength with temperature, as shown in Fig. 9.

The increased thermal shock resistance obtained on inclusion of carbon fibres in glass results partly from the increased work of fracture, so that the material can tolerate a greater amount of mechanical strain generated by a non-uniform temperature distribution, and partly from an increase in thermal conductivity which tends to reduce temperature gradients. The thermal conductivity of Type I carbon fibres is highly anisotropic [5] with a radial value similar to that of the glass matrix, but with a longitudinal value considerably greater. Thus, even a relatively small volume fraction of fibre can be expected to make an appreciable contribution to the thermal conductivity.

7. Conclusions

(1) In hot-pressed carbon fibre composites, thermal stresses resulting from a mismatch in thermal expansion coefficients gave rise to a cracked matrix in magnesia, alumina and soda-lime glass composites, but not in Pyrex or low expansion glass-ceramic.

(2) Matrix cracking was due principally to mismatch with the longitudinal expansion coefficient of the fibres. In the glass and glass-ceramic systems bonding resulted primarily from mechanical keying of the matrix with surface irregularities in the fibres, but the degree of bonding was influenced by mismatch with the radial expansion coefficient of the fibres. In the ceramic systems, bonding resulted primarily from the thermal shrinkage of the matrix on to the fibres.

(3) Keying was more perfect in the case of the glassy matrices because, during hot-pressing, the matrix was in a fluid state.

(4) When the fibres were randomly oriented, all the composites had lower strengths than the matrix materials themselves, but the weakening was less pronounced in glass and glass-ceramic matrices. With partially aligned fibres the strength of borosilicate glass was increased.

(5) Addition of fibres produced substantial increases in the work of fracture for all the materials studied, the relative increase and absolute magnitude being greatest for borosilicate glass. The main contribution is believed to be due to the work done in pulling the ends of fibres out of the fracture surfaces.

(6) The behaviour of the composites at elevated temperatures appeared to be limited by oxidation of the fibres. Composites of zirconia fibres in magnesia, however, showed little change in strength with temperature up to 1500°C. The strength was improved by a preliminary heat treatment at 1600°C which promoted the development of a chemical bond by interdiffusion.

(7) The presence of 10 vol % of carbon fibre markedly increased the thermal shock resistance of borosilicate glass.

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